

Article

Ultrasonic Vibration-Assisted Plasma Cladding of Fe-Cr-C-Based Coatings: Microstructural Regulation and Wear Resistance Enhancement

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Abstract

Fe-Cr-C-based coatings were fabricated on Q690 steel via ultrasonic vibration-assisted plasma cladding at varying ultrasonic powers (0–65 W) with a fixed frequency of 18.5 kHz. The coatings primarily consisted of martensite, retained austenite, and (Cr,Fe)₃C₃ carbides, along with (Cr,Fe,Mo)-B borides along grain boundaries. Increasing ultrasonic power promoted cavitation and acoustic streaming, which refined columnar dendrites, reduced elemental segregation (notably for B and Mo), and increased the fraction of fine equiaxed grains without altering phase composition. As a result, the average microhardness increased from 797.1 to 828.5 HV0.1. The friction coefficient decreased from 0.675 to 0.626, while the wear-track width, wear depth, and wear mass loss decreased from 4.0 mm to 2.5 mm, from 112.5 μm to 32.4 μm, and from 20.40 mg to 4.75 mg, respectively. The wear mechanism shifted from severe adhesive wear to mild abrasive wear. These results demonstrate that increasing ultrasonic vibration power effectively refines the solidification microstructure and significantly improves the hardness and wear resistance of plasma-clad Fe-Cr-C-based coatings

Keywords: ultrasonic vibration; plasma cladding; Fe-Cr-C-based coating; microstructure; wear resistance

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1. Introduction

Excavators are extensively used in construction, mining, and municipal projects. The bucket, acting as the cutting edge, handles gravel and soil, demanding high strength and wear resistance. The bucket teeth and cutting edge are the most wear-prone parts [1–3]. Bucket teeth are replaceable consumables with various wear-resistant materials available, whereas the cutting edge is welded into the bucket structure—its failure often means scrapping the entire bucket; so, enhancing its wear resistance is crucial. Wear-resistant steel plates are commonly used for cutting edges and are classified into high-manganese steels, Hadfield steels, quasi-bainitic steels, and white cast irons based on working conditions and composition. High alloy content improves performance but complicates welding, especially with dissimilar steels like side plates, increasing crack risks. Following the

“two materials for one component” concept, a weldable ordinary alloy steel can satisfy structural needs, while a wear-resistant cladding provides surface protection [4–6]. Q690 high-strength low-alloy steel is widely employed in key load-bearing parts of mining equipment, including excavator cutting edges, bucket teeth, and wear-resistant liners. The steel exhibits desirable mechanical characteristics, such as high strength, satisfactory toughness, and superior weldability, all of which fulfill the performance demands under heavy-load operating conditions.

High-energy beam cladding technologies, including laser cladding and plasma cladding, offer significant advantages over other surface treatment methods for the surface strengthening, repair, and remanufacturing of critical equipment components and have been widely used in marine engineering, the nuclear industry, petrochemicals, and mining equipment [7–9]. Bartkowski et al. [10] prepared Fe/WC composite coatings on low-carbon steel by laser cladding, and the results showed a significant increase in coating microhardness. Zhu et al. [11] prepared AISI 410 martensitic stainless steel coatings reinforced with different TiC particle contents by laser cladding. The coating reinforced with 10 wt.% TiC exhibited higher microhardness and wear resistance, mainly owing to the synergistic effects of microstructural refinement, nanoprecipitation hardening, and second-phase strengthening. Farnia et al. [12] deposited a Co-based alloy coating on a low-carbon steel substrate by laser cladding. The coating was mainly composed of a Co-based solid solution with an FCC matrix, and the substrate underwent a laser-induced phase transformation. Compared with the original substrate, the coating exhibited markedly improved hardness and wear resistance.

Compared with laser cladding, plasma cladding offers controllable and limited substrate melting, with three main advantages: First, it significantly reduces the dilution rate, preventing substrate elements from diluting the coating alloy composition and hard carbides, thereby ensuring high hardness and wear resistance. Second, it minimizes the heat-affected zone, alleviating residual tensile stress and reducing thermal defects such as cracks. Third, it lowers the total heat input, effectively suppressing thermal deformation of large components like the cutting edge [13–15]. Zhu et al. [16] successfully synthesized in situ TiN-Al₂O₃-reinforced CoCrFeNiMn composite coatings by plasma cladding. They found that TiN-Al₂O₃ ceramic reinforcement increased hardness, suppressed adhesive wear, and promoted steady-state friction in the coating. Jun et al. [17] used plasma cladding to deposit a 17Cr2NiSiMo alloy steel coating containing a small amount of Y₂O₃ nanoparticles. The results showed that adding 0.4 wt.% Y₂O₃ nanoparticles significantly improved the wear resistance of the modified coating, reducing wear mass loss by 20.55% and increasing relative wear resistance by 25.87%. Feng et al. [18] prepared an Fe-Co-Mo alloy coating on a 40Cr steel substrate by plasma cladding. The microstructure consisted mainly of an α -Fe matrix and a network of reinforcing phases, and the coating exhibited excellent microhardness and wear resistance.

Because laser and plasma cladding involve rapid melting and solidification, large temperature gradients and residual stresses develop in the coatings after deposition, which can lead to defects such as cracks and pores. Numerous studies have shown that optimizing processing parameters can reduce structural defects in cladding layers to some extent; however, pores, elemental segregation, and uneven distribution of strengthening phases still occur randomly. Therefore, using external-field assistance to control coating microstructure and properties has become an important research topic [19–22].

Through cavitation and acoustic streaming, ultrasonic vibration strongly affects microstructural evolution in the molten pool during solidification, promoting structural homogenization, grain refinement, and residual-stress reduction. Introducing ultrasonic vibration during cladding can improve coating wear and corrosion resistance. Li et al. [23] investigated the effects of high-frequency vibration on the microstructure and properties

of laser-clad WC-Ni coatings and found that vibration treatment produced a more uniform microstructure, thereby increasing microhardness and reducing the coefficient of friction. Zhang et al. [24] fabricated ceramic-reinforced Fe-based composite coatings by ultrasonic vibration-assisted laser cladding. They reported that ultrasonic vibration refined the microstructure, increased the ceramic content, and promoted a more uniform ceramic distribution, mainly because mass and heat transfer in the melt pool were improved. When the laser power was increased to 3000 W, the coating exhibited excellent wear resistance. Zhao et al. [25] prepared a WTaNbMo refractory high-entropy alloy coating on an Inconel 718 superalloy substrate using ultrasonic vibration-assisted laser cladding. The experiments showed that ultrasonic vibration increased the microhardness from 735 HV0.5 to 980 HV0.5. This improvement was attributed to cavitation bubbles generated in the melt pool by ultrasonic vibration; when large dendrites fractured, the resulting fragments acted as new nucleation sites, reducing grain size and thereby lowering the coefficient of friction and wear rate.

Fe-Cr-C wear-resistant coatings are typical in situ carbide-reinforced alloys that have been widely applied in low-stress abrasive wear environments due to their low cost and excellent comprehensive performance [26–29]. Their superior wear resistance primarily originates from the primary carbides $(\text{Cr,Fe})_7\text{C}_3$ precipitated during solidification. The volume fraction, size, distribution, and morphology of $(\text{Cr,Fe})_7\text{C}_3$ directly influence the wear performance of the coatings. In addition, the alloy powder demonstrates good metallurgical compatibility with the Q690 cutting-edge steel substrate, and a low dilution rate is achieved during plasma cladding. Uniform precipitation of hard phases can be obtained without the addition of pre-fabricated ceramic particles, making it ideally suited for the long-term low-stress abrasive wear service conditions of the cutting edge. Therefore, this study employs plasma cladding technology to fabricate Fe-Cr-C coatings on Q690 steel substrate (used for excavator blade plates) while introducing ultrasonic vibration to regulate the solidification behavior of the molten pool and the precipitation morphology of $(\text{Cr,Fe})_7\text{C}_3$. The effects of vibration power on the microstructure and wear resistance of the coatings were systematically investigated.

2. Experimental Materials and Methods

2.1. Coating Preparation

Q690 steel (Suzhou Shoufu Industrial Co., Ltd., Suzhou, China) was selected as the substrate material for the bucket liner in this study, and its nominal composition is listed in Table 1. Before plasma cladding, the substrate surface was cleaned with alcohol and dried to ensure a bright, clean surface free of oil, dust, and other contaminants. FJ-19 Fe-based alloy powder (Tianjin Zhujin Technology Development Co., Ltd., Tianjin, China) was used as the coating powder, and its nominal composition is listed in Table 2. The coatings were prepared using ultrasonic vibration-assisted plasma cladding equipment. Plasma cladding was performed on a PTA-400E4-ST plasma cladding system (Shanghai Duomu Industrial Co., Ltd., Shanghai, China), and ultrasonic vibration was introduced using a JZ1000 auxiliary device (Hangzhou Haoda Ultrasonic Equipment Co., Ltd., Hangzhou, China) —it is composed of an ultrasonic power source, a transducer, and a vibrating platform. The Q690 steel substrate was mounted on the surface of the ultrasonic vibration platform, and the high-frequency vibration was completely conducted from the platform to the whole substrate plate. The cladding parameters are listed in Table 3. Ultrasonic vibration parameters are generally controlled by adjusting the frequency and amplitude. In this study, the vibration amplitude was regulated by varying the ultrasonic power (0, 45, 55, and 65 W), while the frequency was fixed at 18.5 kHz.

Table 1. Chemical composition of Q690 steel (wt.%).

Elements	Fe	C	Si	Cr	B	Mn	S	P
Content	Bal	0.18	0.60	1	0.04	2.00	0.02	0.03

Table 2. Chemical composition of the Fe-based alloy powder (wt.%).

Elements	Fe	C	Si	Cr	B	Mo	Ni
Content	Bal	3.07	2.01	34.09	2.08	3.34	4.34

Table 3. Process parameters for plasma cladding and ultrasonic vibration.

Process Method	Process Parameter	Value
Plasma cladding	Shielding gas	10–15 L/min
	Ion gas	2.0–3.0 L/min
	Powder gas	4.0–5.0 L/min
	Cladding speed	650 mm/min
	First-layer cladding current	165 A
	Second-layer cladding current	150 A
Ultrasonic vibration	Ultrasonic vibration frequency	18.5 kHz
	Ultrasonic vibration power	0 W, 45 W, 55 W, 65 W

2.2. Characterization

Coatings prepared at different ultrasonic vibration powers were sectioned into 15 mm × 10 mm × 12 mm specimens using wire electrical discharge machining. The cladding surfaces and cross-sections were ground to 2000 grit using SiC abrasive papers (Shanghai Aozhong Grinding Technology Co., Ltd., Shanghai, China) on a metallographic polishing machine and then polished to a mirror finish. Phase analysis of the cladding coatings was performed using a Bruker D8 X-ray diffractometer (XRD) (Bruker, Karlsruhe, Germany) over a scanning range of 10° to 90° at a scanning rate of 2°/min. The cross-sectional microstructure, elemental composition, and elemental concentration of the cladding coatings were analyzed using a Quanta-250 scanning electron microscope (SEM) (Hitachi, Tokyo, Japan) equipped with energy-dispersive spectroscopy (EDS), and the worn surface morphology was also observed. An EPMA-8050G (Shimadzu, Kyoto, Japan) electron probe microanalyzer was used to conduct detailed analyses of the microstructure and elemental distribution in different regions of the cladding coating.

2.3. Performance Testing

The microhardness distribution from the coating surface to the substrate was measured using a WILSON VH1102 Vickers (Wilson, Buehler, Lake Bluff, IL, USA) microhardness tester under a load of 100 gf for 15 s, and the results are expressed as HV0.1. Measurements were taken at 150 µm intervals. To ensure accuracy and reliability, three measurements were taken along the same horizontal direction, and the average values were used to plot the hardness distribution curve with error bars. Wear resistance was evaluated using an M-2000 friction (Jinan Jingcheng Testing Technology Co., Ltd., Jinan, China) and wear tester with a ring-on-block configuration and a hardened GCr15 ring (Jiangyin Xingcheng Special Steel Co., Ltd., Jiangyin, China) as the counterface. The ring had a diameter of 40 mm and a hardness of approximately 60 HRC. The test was conducted under a normal load of 120 N at a rotational speed of 200 r/min for 120 min. For each process parameter, three samples were prepared and subjected to wear testing. After testing, abrasive particles on the sample surface were removed using an ultrasonic cleaner and anhy-

drous ethanol, and the samples were dried. The mass change was measured using an electronic balance (BSA224S) (Changzhou Analytical Balance Co., Ltd., Changzhou, China) with a precision of 0.01 mg, and the average value of the three measurements was taken as the mean wear mass loss. A 3D profilometer was used to scan the wear tracks and obtain their three-dimensional profiles.

3. Results and Discussion

3.1. Microstructure of the Coatings

Figure 1 shows the macroscopic cross-sectional morphologies of the coatings prepared at different ultrasonic vibration powers. The coating thickness is approximately 8 mm. The cross-sectional morphologies indicate that the coatings are metallurgically bonded to the substrate. One of the typical characteristics of Fe-Cr-C-based coatings, namely cracking, was also observed. The formation of cracks is determined by the specific compositional system of the coating, which includes the high content of hard and brittle $(\text{Cr,Fe})_7\text{C}_3$ carbides, as well as the high microstructural stress generated during solidification due to the austenite-to-martensite phase transformation. In addition, rapid heating and cooling during plasma cladding produce large temperature gradients and residual tensile stresses. The combined effects of high-alloy-induced microsegregation, internal stress concentration, and the limited plastic deformation ability of the hard and brittle Fe-based phases ultimately lead to crack formation. The crack lengths in the coating regions prepared at different ultrasonic powers (0, 45, 55, and 65 W) are 9.30, 8.76, 8.10 and 7.90 mm, respectively, and the corresponding crack widths are 0.35, 0.21, 0.11 and 0.10 mm, respectively. It can be seen that introducing ultrasonic vibration into the plasma cladding process reduces both the width and length of cracks in the coating, and the degree of improvement increases as the vibration power increases.

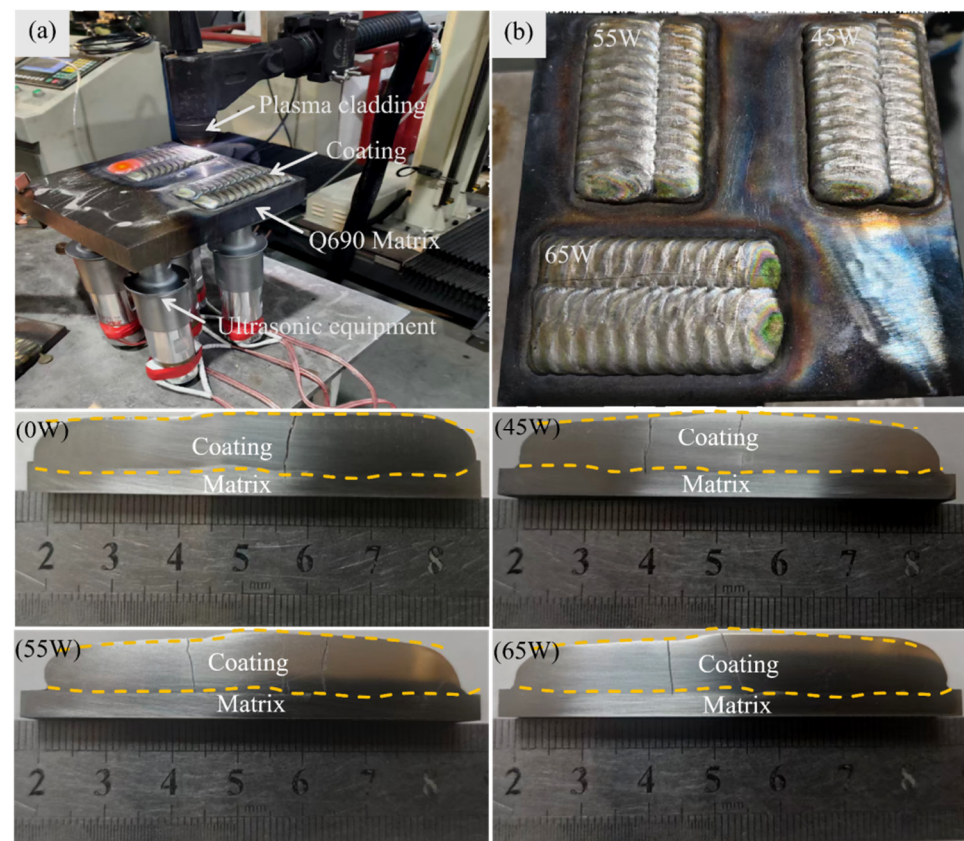


Figure 1. Ultrasonic vibration-assisted plasma cladding process for preparing Fe-Cr-C-based coatings. (a) Experimental setup of ultrasonic-assisted plasma cladding; (b) Macroscopic morphology of the coating.

Figure 2 shows the XRD patterns of the coatings prepared at different ultrasonic vibration powers. As the ultrasonic power increases from 0 to 65 W, the phase composition remains unchanged and is mainly composed of martensite, retained austenite (RA), $(\text{Cr,Fe})_7\text{C}_3$ carbides and $(\text{Cr,Fe,Mo})\text{-B}$ composite borides. The absence of phase transformation indicates that ultrasonic vibration mainly affects melt-pool fluidity and solidification behavior, thereby regulating crystallinity, grain size, and preferred orientation, rather than directly changing the coating phase composition. In addition, the diffraction peaks of the ultrasonically treated coatings are sharper than those of the untreated coating. The enhanced sharpness of the diffraction peaks is primarily attributable to two aspects. On the one hand, ultrasonic vibration promotes the breakup of coarse dendrites via cavitation and the homogenization of solute distribution, which considerably diminishes compositional variations and lattice distortions in the coating, thus effectively relieving the micro-residual strain arising from the solidification process. On the other hand, the refined and uniformly distributed grains, together with the dispersed carbides, mitigate localized stress concentrations at grain boundaries, leading to improved lattice perfection and enhanced crystallinity [30].

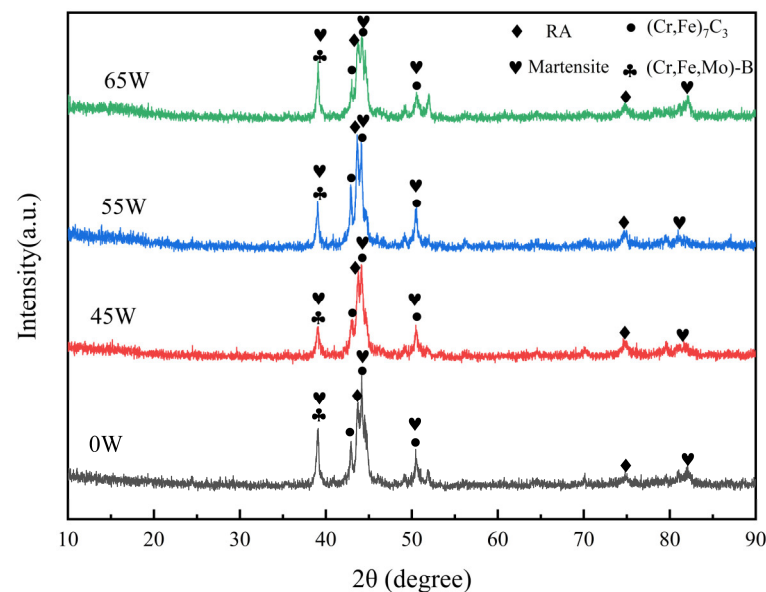


Figure 2. XRD patterns of the coatings prepared at different ultrasonic vibration powers.

Figure 3 shows the microstructures of the coatings prepared at different ultrasonic vibration powers. The coatings exhibit relatively uniform microstructures from the surface to the interior and are mainly composed of a gray matrix phase, columnar dendrites, and equiaxed grains. A clear spatial gradient is observed across the coating thickness: the top region consists mainly of coarse columnar dendrites, the middle region contains both coarse columnar dendrites and equiaxed grains, and the bottom region contains progressively finer columnar dendrites and equiaxed grains.

With increasing ultrasonic power, the columnar dendrites are gradually fragmented and refined, resulting in a significant increase in the number and fraction of equiaxed grains and a marked improvement in microstructural uniformity. ImageJ (1.54f) analysis reveals the average grain sizes in the top (81.03 ± 12.70 , 38.76 ± 7.63 , 38.02 ± 7.42 and 24.66

$\pm 4.40 \mu\text{m}$), middle (33.24 ± 7.15 , 16.86 ± 3.36 , 15.56 ± 3.22 and $12.37 \pm 2.19 \mu\text{m}$), and substrate/coating interface (23.63 ± 4.16 , 23.60 ± 4.09 , 19.86 ± 3.96 and $16.21 \pm 3.35 \mu\text{m}$), respectively. This behavior is attributed to the ultrasonic cavitation generated by ultrasonic vibration. The microstructural refinement mechanism involves dendrite fragmentation, increased nucleation-site density, and reduced compositional segregation, which promote the transformation from coarse columnar grains to fine equiaxed grains. These results indicate that ultrasonic vibration effectively regulates the solidification process of plasma-clad coatings, suppresses columnar grain growth, refines grains, and improves microstructural uniformity, thereby providing a microstructural basis for optimizing coating mechanical properties [31].

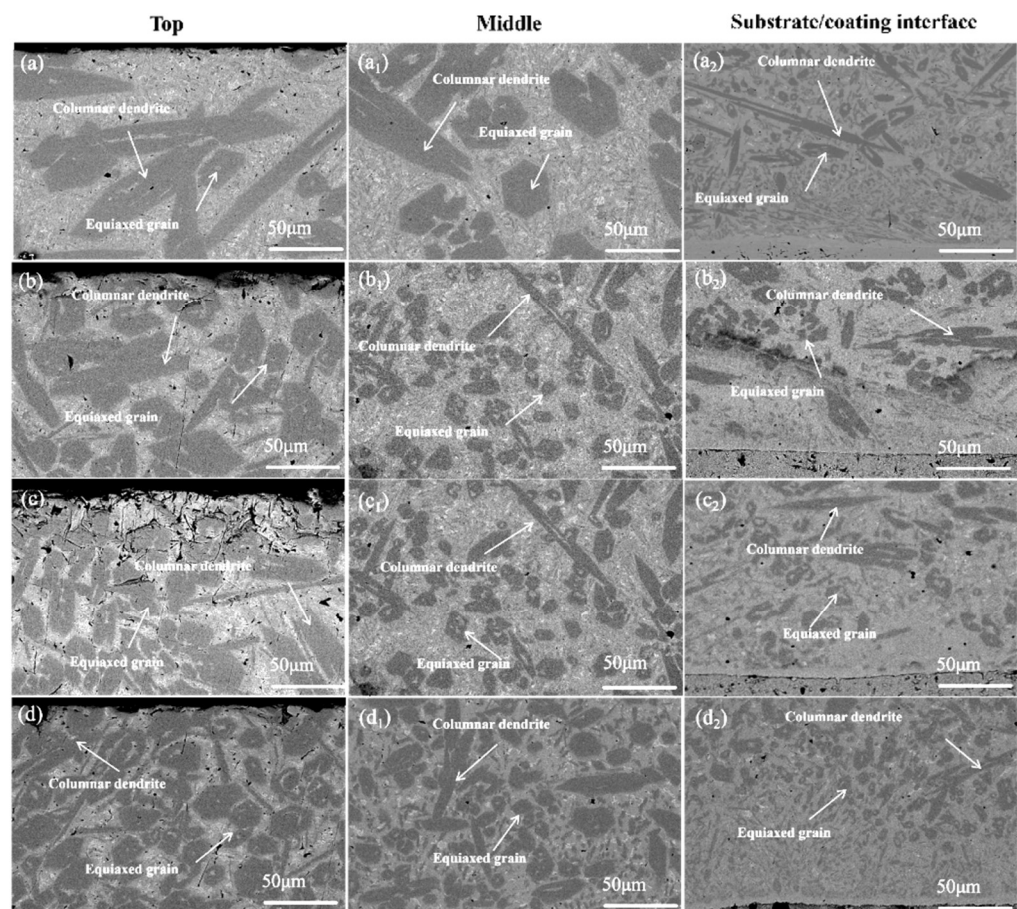


Figure 3. Cross-sectional BSE images of coatings prepared at different ultrasonic vibration powers: (a–a₂) 0 W; (b–b₂) 45 W; (c–c₂) 55 W; (d–d₂) 65 W.

Figure 4 presents the EPMA elemental mapping results for the top cross-sections of the coatings prepared without and with ultrasonic vibration (0 and 65 W). Combined with EDS point analysis (as shown in Table 4), these maps reveal the effect of ultrasonic vibration on coating microstructure and elemental distribution. Without ultrasonic vibration, the coarse primary phase (Point 1) in the top region is strongly enriched in Fe, Cr and C. Based on the XRD pattern in Figure 2, this coarse primary phase is identified as $(\text{Cr,Fe})_7\text{C}_3$. Fe and Ni are mainly distributed in the interdendritic matrix phase (Point 2) as a retained austenite, whereas the grain-boundary phase (Point 3) is a composite borides. After ultrasonic vibration was applied, the $(\text{Cr,Fe})_7\text{C}_3$ phase was refined into uniformly distributed, nearly equiaxed grains (Point 4), and the elemental-distribution differences between the retained austenite (Point 5) and the composite borides (Point 6) were reduced. In other

words, ultrasonic vibration refines the $(\text{Cr,Fe})_7\text{C}_3$ phase, promotes a more uniform distribution of retained austenite, and mitigates the segregation of composite borides.

Table 4. EDS compositional analysis of the marked points in Figure 4 (at.%).

Point	at. %						
	Fe	C	Cr	Mo	Ni	B	Si
1	24.25	23.45	50.64	1.04	0.63	-	-
2	62.91	6.03	11.87	0.60	8.97	-	9.62
3	32.76	5.45	14.64	13.39	2.99	26.46	4.31
4	22.18	24.25	46.70	1.05	0.62	5.20	-
5	61.89	13.03	16.73	4.57	1.65	-	2.14
6	31.12	2.44	17.37	14.69	2.14	29.08	3.15

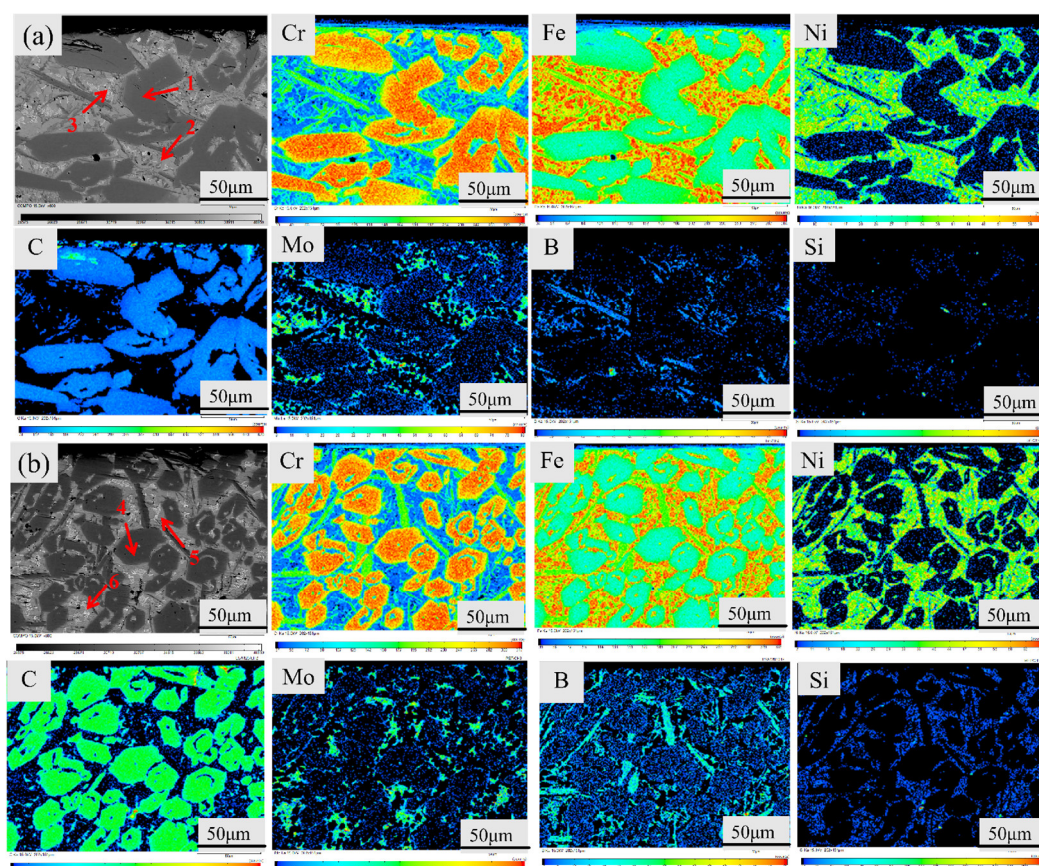


Figure 4. EPMA maps of the top cross-sections of coatings prepared without and with ultrasonic vibration: (a) 0 W; (b) 65 W.

Figure 5 presents the EPMA elemental maps and EDS point-analysis results for the bottom cross-sections of the coatings prepared without and with ultrasonic vibration (0 and 65 W). Compared with the top region of the coating (Figure 4), the bottom region exhibits a significantly finer microstructure, further demonstrating the effect of ultrasonic vibration on microstructure and elemental distribution. As shown in Table 5, without ultrasonic vibration, the bottom of the coating contains slender $(\text{Cr,Fe})_7\text{C}_3$ phase regions (Point 1); EDS point analysis indicates that the Cr and C contents are as high as 44.12 at.% and 25.01 at.%, respectively. Fe and Ni are distributed in the interdendritic matrix as retained austenite (Point 2), while composite borides are randomly distributed in the interdendritic matrix as grain-boundary phases (Point 3).

After ultrasonic vibration was applied, the $(\text{Cr,Fe})_7\text{C}_3$ phase at the bottom was also fragmented into uniformly distributed, nearly equiaxed grains, and the distribution of Fe and Ni in the retained austenite phase became more uniform. EDS point analysis showed that the compositional differences among the $(\text{Cr,Fe})_7\text{C}_3$ phase (Point 4), the retained austenite (Point 5), and the composite borides (Point 6) were significantly reduced, and grain-boundary segregation of coating elements was markedly suppressed. This indicates that ultrasonic vibration, through cavitation and acoustic-streaming effects, effectively fragments the coarse $(\text{Cr,Fe})_7\text{C}_3$ phase, promotes homogeneous elemental diffusion in the molten pool, improves compositional uniformity, and enables synergistic regulation of microstructure and composition [32,33].

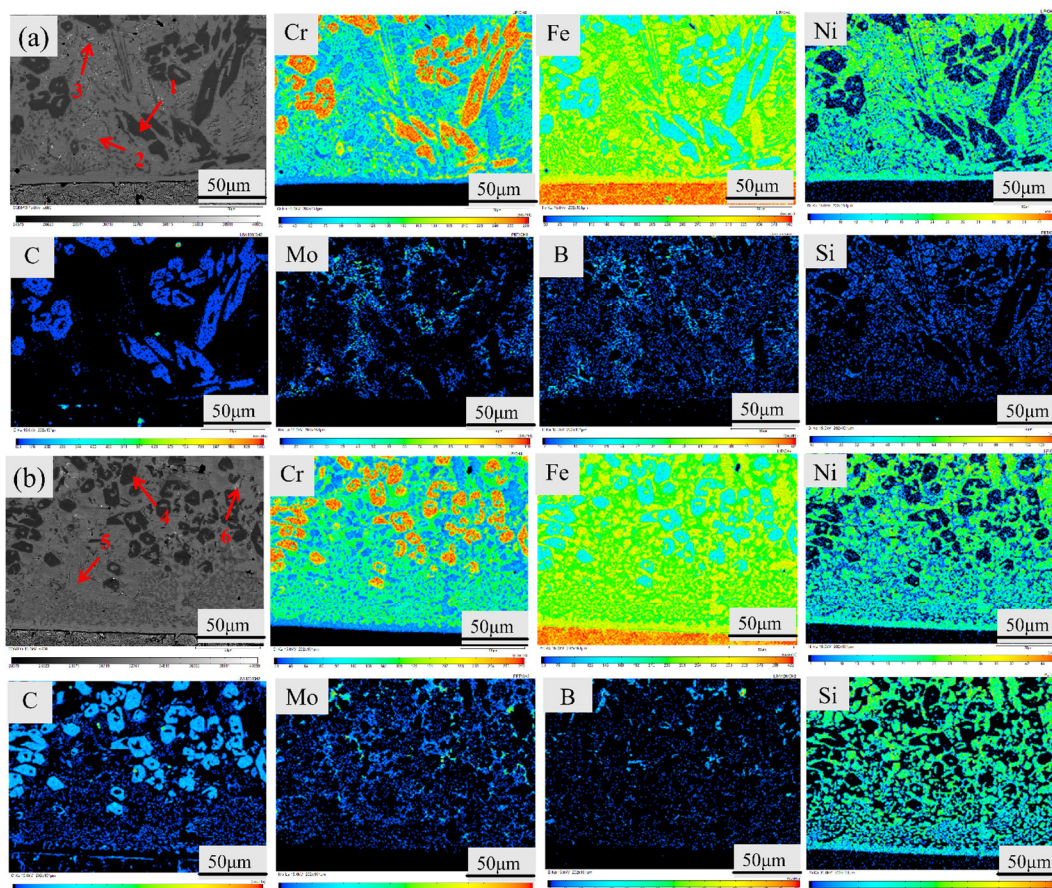


Figure 5. EPMA maps of the bottom cross-sections of coatings prepared without and with ultrasonic vibration: (a) 0 W; (b) 65 W.

Table 5. EDS compositional analysis of the marked points in Figure 5 (at.%).

Point	at. %						
	Fe	C	Cr	Mo	Ni	B	Si
1	24.60	25.01	44.12	0.86	-	5.41	-
2	72.15	5.02	10.68	0.51	4.92	-	6.72
3	35.30	1.19	30.66	0.58	0.44	31.69	0.15
4	35.55	28.23	33.08	0.58	0.45	2.11	-
5	71.80	4.77	10.34	0.41	5.65	-	7.03
6	42.61	3.95	12.36	10.05	2.69	25.82	2.53

3.2. Hardness Analysis

Figure 6 shows the hardness distribution and average hardness values of the coatings prepared at different ultrasonic vibration powers. As shown in Figure 6a, the cross-section can be divided into three regions according to the hardness distribution: the coating zone, the heat-affected zone (HAZ), and the substrate. The coating exhibits the highest microhardness, followed by the HAZ, whereas the substrate has the lowest microhardness. As shown in Figure 6b, the average coating hardness increases gradually with increasing ultrasonic power. At ultrasonic powers of 0, 45, 55, and 65 W, the average microhardness values are 797.1, 811.0, 817.4, and 828.5 HV_{0.1}, respectively. The relatively low hardness of the coating prepared without ultrasonic vibration is attributed to its coarse grains. Ultrasonic cavitation and acoustic streaming broke up the coarse columnar dendrites, leading to a significant increase in grain boundary density. Meanwhile, the coarse blocky (Cr,Fe)₇C₃ carbides were refined into fine and homogeneously distributed particles. The synergistic strengthening from both grain refinement and hard carbide dispersion contributed to a further improvement in the microhardness of the coating.

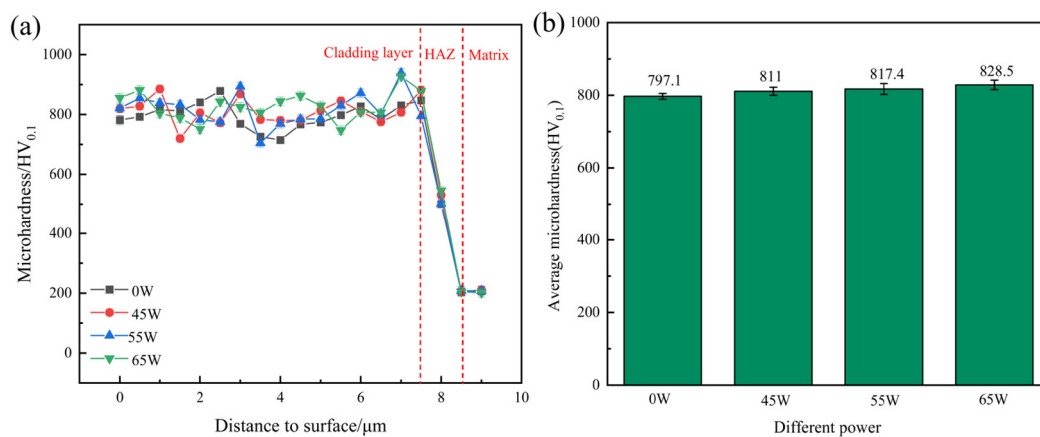


Figure 6. Microhardness distribution and average hardness of coatings prepared at different ultrasonic vibration powers. (a) Microhardness distribution, (b) Average hardness.

3.3. Wear Resistance

Figure 7 shows the friction coefficient curves of the coatings prepared at different ultrasonic vibration powers. As shown in Figure 7a, all coatings undergo a running-in stage followed by a steady-state wear stage. During the later stage of steady-state wear, wear debris affects the smoothness and contact area of the friction interface, causing the coefficient of friction to increase continuously and fluctuate. Figure 7b shows that the average friction coefficients of the coatings prepared at 0, 45, 55, and 65 W are 0.675, 0.669, 0.651, and 0.626, respectively. The average coefficient of friction decreases gradually with increasing ultrasonic vibration power. Compared with the coating prepared without ultrasonic vibration, the coefficient of friction decreases by 7.3% at 65 W, indicating that ultrasonic vibration effectively suppresses wear between the coating surface and the counterface and thereby improves wear resistance.

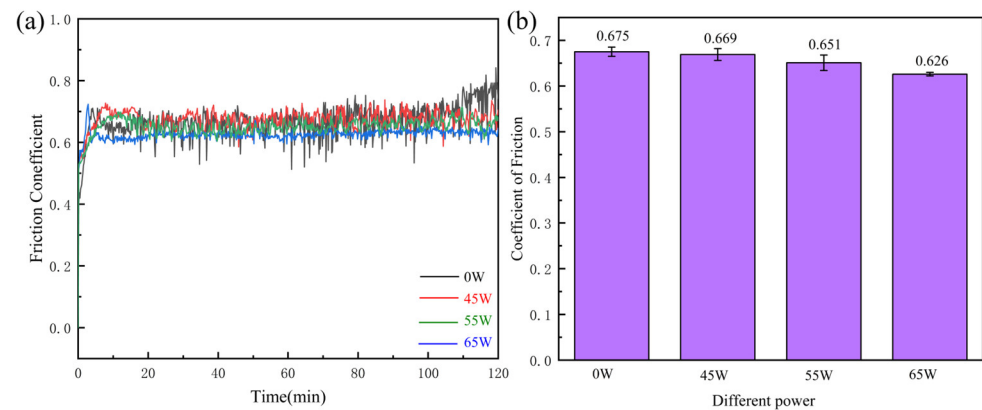


Figure 7. Friction coefficients of coatings prepared at different ultrasonic vibration powers. (a) Friction coefficients, (b) Average friction coefficients.

Figure 8 shows the wear morphologies of the coatings prepared at different ultrasonic vibration powers under identical wear-test conditions. As shown in Figure 8a–d, the wear-track widths are 4.0, 3.3, 3.0, and 2.5 mm for the coatings prepared at 0, 45, 55, and 65 W, respectively. The macroscopic morphologies indicate that wear is most severe at 0 W; with increasing ultrasonic power, wear damage is reduced and coating wear resistance improves. To further compare the wear severity among the coatings, Figure 8(a1–d1) presents the corresponding 3D wear-track profiles. As the ultrasonic power increases, the width and depth of the wear track gradually decrease. Combined with the wear-depth and mass-loss data in Figure 9, these results show that increasing the ultrasonic power from 0 to 65 W decreases the wear depth from 112.5 to 32.4 μm and the wear mass loss from 20.40 to 4.75 mg, indicating a significant improvement in wear resistance.

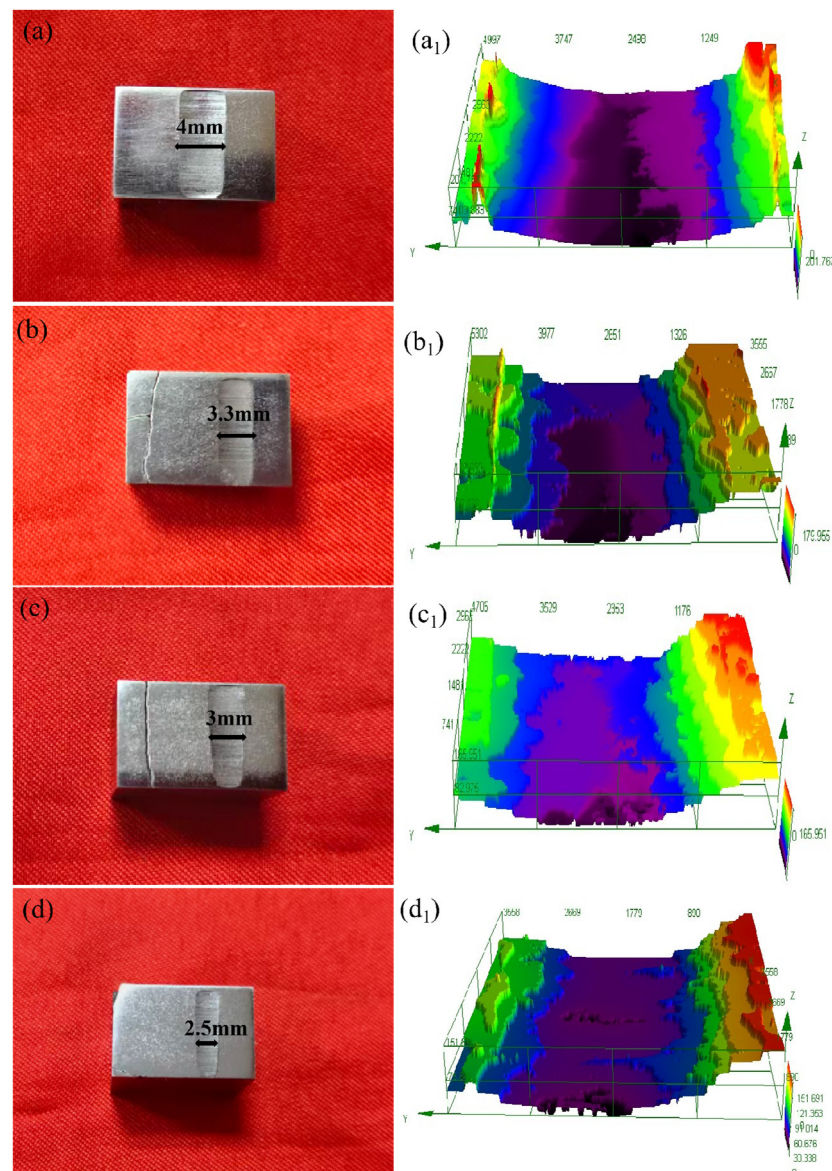


Figure 8. Macroscopic morphologies and 3D profiles of coating wear tracks at different ultrasonic vibration powers: (a,a₁) 0 W; (b,b₁) 45 W; (c,c₁) 55 W; (d,d₁) 65 W.

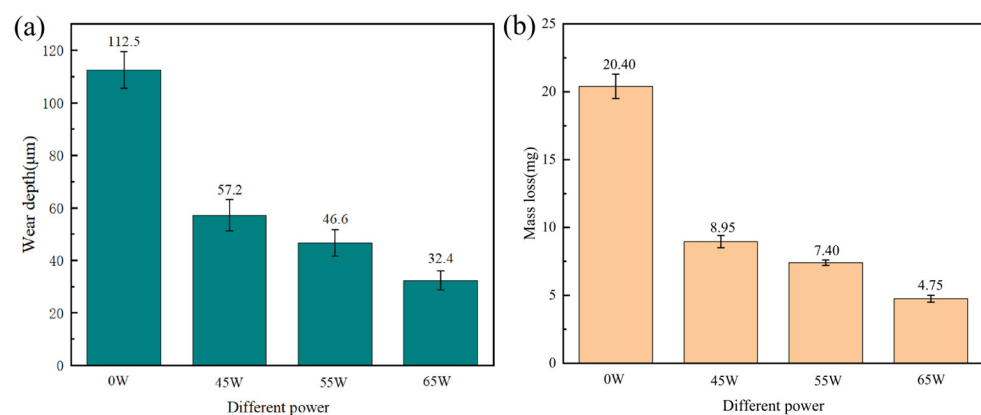


Figure 9. Wear depth and mass loss of coatings prepared at different ultrasonic vibration powers. (a) Wear depth, (b) Mass loss.

Figure 10 shows SEM images of the worn surfaces of coatings prepared at different ultrasonic vibration powers. As shown in Figure 10a, when no ultrasonic vibration was

applied, the worn surface exhibited deep grooves, spalling, delamination, and abundant debris, indicating severe adhesive wear. As the ultrasonic power increases (Figure 10b,c), only a small amount of debris is observed on the worn coating surfaces, and spalling is significantly reduced; in this case, the dominant wear mechanism changes to abrasive wear. At an ultrasonic power of 65 W (Figure 10d), the coating surface is relatively smooth with only a few pits, indicating mild abrasive wear. Based on the aforementioned microstructural characteristics of the coatings, it can be concluded that in the absence of ultrasonic treatment, coarse dendrites and elemental segregation lead to localized brittle cracking, spalling, and adhesive wear. At 65 W, the refinement and uniform distribution of $(\text{Cr,Fe})_7\text{C}_3$ result in more homogeneous load-bearing capacity and reduced localized stress concentration, thereby transforming the wear mechanism from severe adhesive wear/spalling to mild abrasive wear.

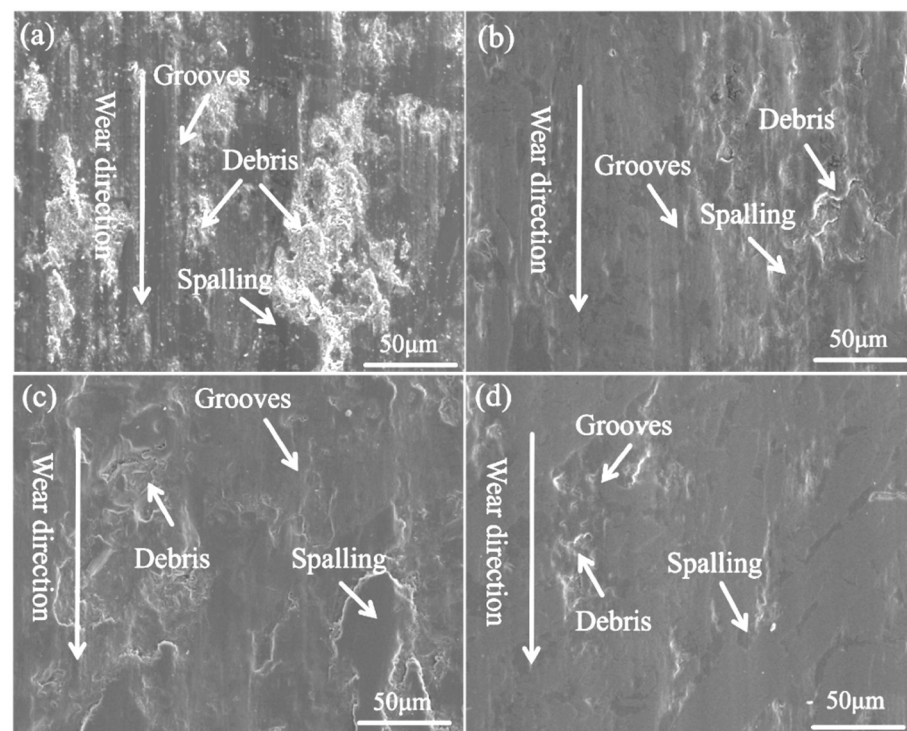


Figure 10. BSE images of worn coating surfaces prepared at different ultrasonic vibration powers: (a) 0 W; (b) 45 W; (c) 55 W; (d) 65 W.

Figure 11 illustrates the mechanism by which ultrasonic vibration regulates solidification microstructure evolution in coatings prepared at different ultrasonic powers (0–65 W). Without ultrasonic treatment (Figure 11a), forced convection is absent in the melt pool, and columnar crystals preferentially grow from the bottom of the melt pool along the temperature gradient. Consequently, the microstructure contains a high fraction of columnar crystals, few equiaxed crystals, and coarse grains. After ultrasonic vibration is introduced (Figure 11b–d), ultrasonic cavitation and acoustic streaming significantly alter melt-pool solidification behavior. On the one hand, the localized high pressure and microjets generated by cavitation-bubble collapse fragment the growing dendrite arms of columnar crystals and provide numerous free nucleation sites for equiaxed grains. On the other hand, acoustic streaming promotes forced convection in the melt pool, homogenizes temperature and solute distributions, suppresses directional columnar-crystal growth, and facilitates the transition from columnar to equiaxed crystals [30].

As the ultrasonic power increases from 45 to 65 W, the ultrasonic energy density increases, leading to stronger dendrite fragmentation and grain refinement. The fraction of

equiaxed grains in the cladding layer gradually increases, the grain size becomes more uniform, and columnar-grain growth is more strongly suppressed. These results demonstrate that increasing ultrasonic power effectively regulates the solidification microstructure of the cladding layer by enhancing cavitation and acoustic-streaming effects, thereby achieving grain refinement and microstructural homogenization and ultimately improving coating hardness and wear resistance.

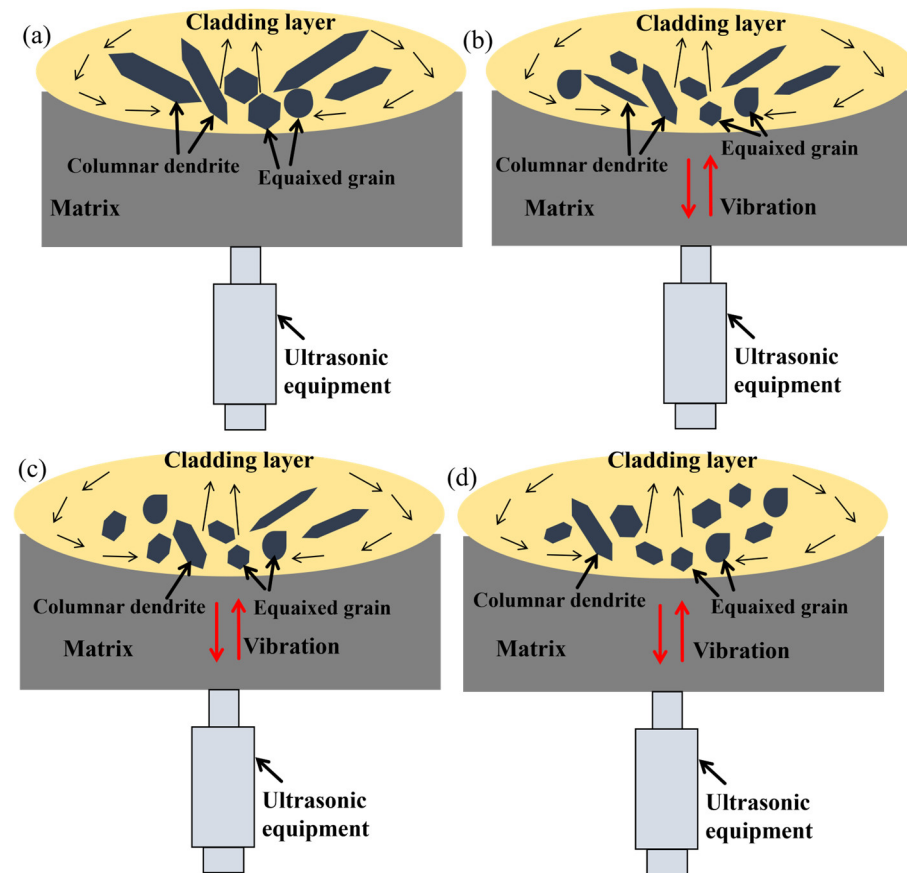


Figure 11. Schematic diagram illustrating the mechanism by which ultrasonic power affects coating microstructure: (a) 0 W; (b) 45 W; (c) 55 W; (d) 65 W.

4. Conclusions

(1) Introduction of ultrasonic vibration did not change the phase composition of the plasma cladded Fe-Cr-C-based coatings, which was predominantly composed of martensite and a small amount of retained austenite, with extensive precipitation of primary and eutectic $(\text{Cr,Fe})_7\text{C}_3$ carbides, as well as composite borides of $(\text{Cr,Fe,Mo})_7\text{B}$ distributed along grain boundaries. Ultrasonic vibration refines the $(\text{Cr,Fe})_7\text{C}_3$ phase, promotes a more uniform distribution of martensite matrix, and mitigates the segregation of composite borides.

(2) Microhardness analysis showed that as the ultrasonic vibration power increased from 0 to 65 W, the average microhardness of the coating increased from 797.1 to 828.5 HV0.1, confirming a significant grain refinement effect.

(3) Friction and wear tests under identical conditions showed that increasing ultrasonic vibration power reduced the average coefficient of friction from 0.675 to 0.626, decreased the wear-track width from 4.0 to 2.5 mm, and reduced the wear mass loss from 20.40 to 4.75 mg. The dominant wear mechanism changed from severe adhesive wear to mild abrasive wear.

This study demonstrates that ultrasonic-assisted plasma cladding can improve the microstructure and wear resistance of Fe-Cr-C coatings on Q690 steel, which is applicable to the repair of heavy-load wear-resistant parts in mining machinery with low equipment retrofitting cost and favorable mass production potential. Future research will focus on multi-external-field coupled cladding, alloy powder composition optimization and pilot tests on large-sized components to establish standardized procedures and advance the industrial mass application of this wear-resistance-strengthening technology.

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